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The Metallurgist.

Supplement to **The Engineer**, *February, 1935.*

The Use of Metals.

A CONSIDERATION of the part which metals have played in the development of civilisation is always an intriguing task, and the subject was dealt with in an extremely interesting manner by Mr. A. J. Grant in his Presidential Address given to the Sheffield Society of Engineers and Metallurgists on January 21st. As a highly trained engineer and industrialist, it was not surprising that he should emphasise the factors governing the economic aspects of the matter, and one of his most interesting observations was to the effect that where a metal is demanded in great tonnage its cost of production falls to economically low values, through the inevitable production in large efficient manufacturing units. Iron produced in modern blast-furnace plants was the outstanding metal instanced. Occurring in the earth's crust in less quantity than aluminium, iron in its cheaper forms is available at such relatively trivial prices as is likely to assure a continuance of the Iron Age (or had we better say Steel Age ?). Metals produced in small quantities, even by simple and efficient processes, were at a disadvantage with metals produced in great quantities by efficient though complex processes.

It was easy to conceive and to establish, say, for a new metal or alloy, a simple and efficient process in the laboratory, but the world could not take advantage of the achievement until such metal was manufactured on an extensive scale, and the efficiency of the process was confirmed in low costs of production. To manufacture on an extensive scale demands a large market, and that can only be achieved either by the industrialist putting the metal in such a case before the public at a price reasonably near that which might be anticipated as practicable on the combined evidence and judgment available, or by painfully building up through decades or generations an increasing consumption of the new metal or alloy at economic values coincident with an adequate return at the costs of production existing in the earlier and attenuated state of the development of the process. Thus is needed the enlightened judgment and outlook of the industrialist backed by adequate finance and an efficient commercial machine, if metals and alloys are in a reasonable time to attain the position to which they are entitled by their intrinsic properties and potential ultimate availability.

Mr. Grant instanced another very interesting comparison of two metals, in this case produced in similar quantities, *i.e.*, cobalt and gold. In each case the world's output was of the order of 800 tons for last year. The gold, great in demand and insufficient in availability, realised approximately £240,000 per ton ; the cobalt, insufficiently in demand, but available in excess, could be obtained at little above £400 per ton. Therefore it is clear that the market value does not necessarily always outweigh the intrinsic characteristics which determine the quantity in which a metal is used.

After a very interesting survey from the archæological down through the historical period, the address dealt with the possibilities of the future. In this connection it was stated, and this cannot be emphasised too strongly, that " the engineering conception based on increasing knowledge of physical laws, can, at times, see the mechanical progress possible, *provided a suitable metal is available having the necessary mechanical and physical properties*," *i.e.*, the more completely the engineer and physicist disclose their requirements and hopes, thus giving direction of thought to the metallurgical investigator, the greater the chance that their needs will be met. Bearing upon this, the importance of a complete knowledge of the physical constants of the many diverse alloys was emphasised, and interesting cases cited where such knowledge had given simple solutions to apparently extremely difficult problems.

In conclusion, Mr. Grant indicated his belief that owing to its vast tonnage and the efficient manner in which it was produced, iron would remain the basic metal of our civilisation for a long period of time. Particularly was he of this opinion in view of the widely dissimilar characteristics which could be introduced as a result of alloying other elements with it. After a brief review of the great achievements in this direction, the view was expressed that amongst the many combinations possible in the ferrous systems as yet unexplored many outstanding alloys yet remain to be discovered.

Such may, indeed, be the case, and the address will no doubt do much further to inspire the many investigators at work in this field. " The Use of Metals " was a fortunate subject for an address which cannot fail to assist in the closer collaboration between the engineer and the metallurgist, and in encouraging the latter in the exploitation of the possibilities of his metals.

Applications of Zinc Alloys.

SOME possible applications of zinc alloys to domestic gas and water distribution plant are suggested by R. Cazaud and H. Pétot in an article in *Le Génie Civil* for July 14th, 1934 (Vol. 105, pp. 34–37). After describing some of the disadvantages of zinc coatings on iron, particularly the danger of electrolytic attack when the zinc wears off, they urge the desirability

of developing a zinc alloy with high resistance to corrosion (greater even than that of zinc itself), together with sufficient mechanical strength to dispense with the necessity of support by a stronger underlying metal. They claim, after prolonged trials, to have found such a material in the 5 to 20 per cent. aluminium-zinc alloys, prepared from electrolytic zinc and pure (99·5 per cent.) aluminium, and improved by the addition of a small quantity of a third metal. These special alloys have been patented and have received the name of " totalin " (*total*ement *in*altérable).

On prolonged exposure to air, zinc is slowly covered with a thin protective layer of hydrated carbonate which is impermeable to, and insoluble in, water. In ordinary water containing calcium salts a similar protective layer is formed. The harder the water the more quickly is the protective film produced. Alloys are required which show the maximum possible rapidity of film formation and a more powerful auto-protection than zinc itself. This high degree of auto-protection is said to be a characteristic of the " totalin " alloys. They have proved to be resistant to water from many different sources, even to water strongly charged with chlorine or ozone.

The small addition of a third metal has the effect of refining the grain size and improving the mechanical properties. There is an optimum amount for each added element. For antimony it is 0·5 per cent.; increase of antimony to 1 or 2 per cent. further refines the structure, but the mechanical properties are impaired. One metal may be added to the alloy, or several simultaneously. Magnesium, antimony, and titanium are most beneficial ; cadmium, tin, and bismuth are least effective. The addition of a metal which refines the grain size suppresses the phenomenon of ageing which detrimentally affects many zinc alloys ; the alloys become in some way stabilised.

The mechanical properties of the alloys in the form of tube are as follows :—

Dimensions of tube, mm.	Al, per cent.	Elastic limit.	Tensile strength.	Elongation, per cent.
		Tons per sq. inch.		
33–38	5	16·5	19·4	23
10–15	10	12·1	20·3	25
26–30	10	16·8	20·2	35

The tensile strength can be raised by cold-drawing, with a corresponding reduction in elongation. Hydraulic pressure tests gave the following results :—

Diameter.		Pressure, tons per sq. in.	Remarks.
Outside.	Inside.		
12·8 mm.	10·0 mm.	0·63	Without bursting or deformation
30·0 mm.	26·0 mm.	1·6	

Diameter.		Bursting pressure, tons per sq. in.
Outside.	Inside.	
39 mm.	36 mm.	1·28
15 mm.	10 mm.	6·5

The toxicity of the water which had passed through these pipes was estimated to have been not more than one-eighth that of water in similar contact with copper pipes.

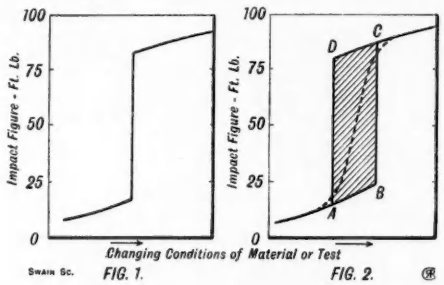
Irregular Impact Figures.*

By R. H. GREAVES, D.Sc.

IT sometimes happens that adjacent specimens of the same steel (or adjacent notches in the same test piece) show a large variation in notched-bar impact figure. This has been attributed, from time to time, either to gross lack of homogeneity in the material or to unavoidable variation in the radius of the notch, differences in microstructure of the material at the root of the notch, or irregular energy losses arising from the design of the testing machine. It should be recognised, however, that within certain ranges of tensile strength or (for a given tensile strength) at certain sizes of section treated the conditions are present in nearly every steel for the occurrence of irregular or scattered values of the impact figure.

In the breaking of an impact test piece two modes of fracture may be involved : (*a*) failure by shear with considerable deformation ; and (*b*) failure by separation under three dimensional tensile stress with little or no deformation. The evidence for and consequences of this are dealt with in a series of papers on " The Notched Bar Impact Test," presented at the Zürich Congress of the International Association for Testing Materials, 1931.

The change over from one type of fracture to the other with changing conditions of material (*e.g.*, composition, tensile strength, tempering temperature, mass treated, &c.) or of test (*e.g.*, form of notch, test piece, speed of impact, temperature, &c.) is not perfectly sudden, as in Fig. 1. The possibility of the occurrence of each form of failure persists over a certain range, in which the two types of fracture are mixed to a varying degree and in which the test



SWAIN Sc. FIG. 1. FIG. 2.

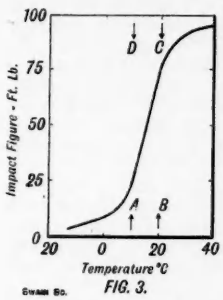
pieces therefore give scattered results. This is shown diagrammatically in Fig. 2. In the vicinity of A D there is a bigger proportion of low results, and near B C a bigger proportion of high results, so that on the whole the average effect is shown by a steeply rising impact figure curve.

The impact figure of a material shows scatter if the impact figure/temperature curve rises steeply in the neighbourhood of atmospheric temperature. A steel with the impact figure/temperature curve shown in Fig. 3 would give irregular results between A D and B C, *i.e.*, if tested between 10 and 20 deg. Cent., but consistent impact figures if tested above 25 or below 3 deg. Cent.

In an ordinary heat-treated steel scattered results are obtained in the region of rapid rise, if the impact figure rises rapidly with fall in tensile strength, but

* Communication from the Research Department, Woolwich.

not if the impact figure curve is reaching its maximum. For example, a manganese-molybdenum steel (carbon 0·38, manganese 1·42, molybdenum 0·23 per cent.), when heat-treated in the form of 2½in. diameter bar to a tensile strength of 50 tons per square inch, gave impact figures varying from 50 to 81 ft.-lb., but when re-tempered to give a tensile strength of 45 tons per square inch it consistently gave an impact figure within 1 or 2 ft.-lb. of 85. Thus in Fig. 4 steel treated to a tensile strength Q gives a uniformly high impact figure, but with a tensile strength P it may give



SWAIN SC. FIG. 3.

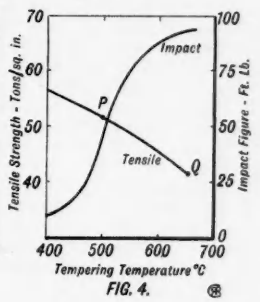
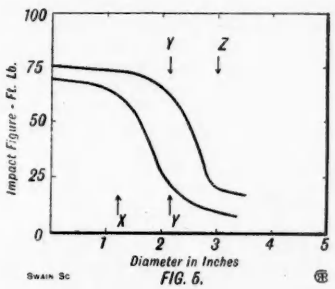


FIG. 4.

irregular results. The remedy is to re-temper steel in the condition P and bring the tensile strength down. If this is not practicable, because the tensile strength is already at the bottom of the particular specification which applies, a steel of slightly higher alloy content must be used.

The effect of mass is also important, certain dimensions being critical for a particular tensile strength in a given composition. Thus in Fig. 5, if a tensile strength of, say, 50 tons per square inch is required,



SWAIN SC FIG. 5.

the conditions as regards impact figure become critical at a diameter X, and scattered results occur with pieces of diameters X to Y. (It is here assumed that the required tensile strength can always be obtained, in the sizes dealt with, by suitable choice of tempering temperature; mass effect shows in impact figure and in yield ratio before it shows in tensile strength.) If the same steel is used at a tensile strength of, say, 42 tons per square inch the region of scatter is raised to a higher diameter, and irregular impact figures now occur in the range of diameters Y to Z. Such a diagram as Fig. 5 is typical of a carbon-manganese steel or a nickel steel of very low nickel content. A 3½ per cent. nickel steel shows a similar effect in larger sections. For a manganese-molybdenum steel of the type referred to above, heat treated in different sections to give a tensile strength of 55 tons per square inch, the steep part of the impact figure diameter curve is at about 3in., and the variation of impact figure is as follows :—

Diameter.	Range of impact figure.	Average.
1in.	58 to 61	60 ft.-lb.
2in.	57 to 59	58 ft.-lb.
3in.	25 to 44	33 ft.-lb.
4½in.	27 to 32	30 ft.-lb.
6in.	25 to 28	26 ft.-lb.

Sections at which the scatter effect is encountered are on the steep part of the impact figure/diameter curve, and the position of the steep part, for a given tensile strength, is in turn related to the composition of the steel.

It will be clear from this discussion that irregular impact figures do not necessarily indicate any appreciable lack of homogeneity in the material, but are consequent on the operation, to various degrees, of the alternative possible modes of failure of the material under test. Over a certain critical range of conditions—very slight accidental differences are sufficient to turn the balance in favour of one or other of these modes of failure with an exaggerated effect on the energy absorbed.

The views expressed above are based on general experience of the heat treatment of a variety of alloy steels, and not on any complete series of tests made on the same steels throughout, with the object of confirming the form of all the diagrams given. It is considered, however, that these general statements may serve as a reliable guide in the consideration of methods of avoiding irregular impact figures and developing high impact steels generally.

Creep at Elevated Temperatures.

SINCE 1922, when Dickenson's[1] pioneer work on the creep of steels at elevated temperatures appeared there has been an uninterrupted spate of papers, many of which are reviewed in Tapsell's[2] well-known book, describing the properties of metals and alloys, particularly steels, at elevated temperatures. The writer has before him papers by Körber,[3] Dustin[4] and Ranque and Henry[5] which were communicated to the Third Congress of Industrial Heating and which have appeared in recent numbers of *Revue de Métallurgie*, and also a paper by McVetty[6] on the interpretation of creep tests. In the course of the following article an attempt will be made to deal briefly with the foundations and recent developments of the subject, especial attention being given to the above-mentioned papers.

The study of metals under prolonged stress and at elevated temperatures is of comparative antiquity, although it is only in recent years that investigations have been carried on with the present intensiveness, which arises as the result of demands for new information and new materials, coming in the main from the power industry. The demand of the power producer, who, in the interests of economy, wishes to increase steam pressures and/or temperatures, is for materials which under these conditions "should permit of the same unit stresses as materials at present in use below 400 deg. Cent."[7] A smaller but no less insistent demand is for materials capable of withstanding successfully the high service temperatures within

industrial furnaces. Although resistance to oxidation and other factors are undoubtedly of great importance in such applications, it is now universally agreed that resistance to deformation and to rupture under service stress is the most important criterion.

It has long been known that the time factor is of great importance in determining the response of metals to tensile stress. This was first conclusively shown to apply to tests at room temperature by Maitland[8] in his experiments upon unhardened gun steel, in which he tested the material :

A, at a normal rate of strain in a testing machine.
B, by allowing dead-loaded test pieces to fall vertically and arresting the upper end by stops.
C, by firing a charge of guncotton between plugs into which were screwed the test piece and which themselves were a good sliding fit within a strong tube.

By these three methods he obtained elongation values of A, 27 per cent.; B, 47 per cent.; C, 62 per cent. The method of experiment did not allow of measurement of changes in strength.

In early experiments upon the effect of temperature on the mechanical properties of metals and in the determination at room temperature of the strength of soft metals, such as lead, little attention was paid to the rate of deformation. Indeed, values for the tensile strength of metals at elevated temperatures—for example, silver at 200 deg. Cent.[9]—have been given to five figures, yet unaccompanied by any reference to the duration of test. It is to be regretted that the description of the properties of metals at elevated temperatures in an incomplete and misleading way still occurs to-day, despite our present knowledge of "creep."

The word "creep" was probably first employed in its present significance by Muir[10] in his classical study of the recovery of steel from overstrain, where it is used to describe flow under constant load of iron stressed at room temperature above its yield point. Muir also employed the term "creep-back" to signify dimensional recovery after overstrain.

The earliest study of metals at elevated temperatures under constant load were those of Rosenhain and Humfrey[11] upon iron, which were carried out in an attempt to secure practical support for Beilby's[12] amorphous cement theory as modified by Rosenhain and Ewen.[13] Rosenhain and Humfrey put forward a power law connecting rate of strain with applied stress, and as a result of their extremely painstaking microscopical work, were able to differentiate between the mechanisms of flow and fracture in iron at low temperatures or fast rates of loading, and those at higher temperatures under slow strain. The work of Rosenhain and Ewen[14] upon non-ferrous metals was carried out by raising the temperature of a specimen while it was subjected to load, a method later favoured by Ingall.[15] Such tests cannot be regarded as true creep tests. The micrographical results, however, confirmed those of the constant load, constant temperature tests on iron, and proved that at low temperatures failure was by crystalline slip and resulted in transcrystalline fracture, whilst at high temperatures, provided the rate of strain was not excessive, there was little or no development of slip bands and the fracture was intercrystalline with little general elongation or necking. Hanson[16] has confirmed these results in the case of aluminium, and has shown that at the atmospheric temperature single crystals are probably stronger than polycrystalline aggregates. According to Jenkins,[17] an investigation which is at present in progress as to the mechanism of failure of iron and steel when subjected to prolonged stress in vacuo at elevated temperatures has also confirmed Rosenhain's earlier study.

In this connection it is interesting to observe that Pearson,[18] in his recent study of the viscous properties of extruded lead-tin and bismuth-tin eutectic alloys, has found that the rate of flow of these materials under constant stress is extremely sensitive to the duration of ageing following extrusion and preceding test. Ageing of these alloys has the effect of producing grain growth and the alloys with the largest grain size have the highest creep resistance or lowest rate of flow under a given stress. The crystals do not exhibit slip bands or other evidence of internal movement; hence flow is presumably confined to the grain boundaries, yet the total extension is great and fractures are not of the brittle type associated by Rosenhain with intercrystalline failure.

Rosenhain's hypothesis demands that one considers a polycrystalline aggregate as consisting of, first, the crystals themselves, and secondly, intercrystalline matter having the properties of a super-cooled liquid, the latter being stronger at low temperatures, but its strength falling off more rapidly with increase in temperature, so that at temperatures approaching the melting point it is weaker than the crystals. As strength is that of the weakest link, the strength of the aggregate would be that of the crystals at the lower temperatures, but above the temperature of intersection of the curves, designated the equi-cohesive temperature, would follow that of the intercrystalline material. Such a tensile strength curve as Rosenhain postulated had already been obtained by Bengough,[19] and has since been found repeatedly; but it must be pointed out that creep-stress/temperature curves of this two-branch type have never been described, and indeed life/temperature curves at constant stress are usually of a form which Dickenson[1] describes as exponential and Kerr[20] as hyperbolic.

Ranque and Henry[5] attach particular importance to the equi-cohesive temperature and consider that above this temperature grain size is the controlling factor in determining creep-rate under a given load. They point out that below this temperature materials may be improved in their resistance to creep by the precipitation during test of a constituent at first entirely held in solution. This they have demonstrated in the case of steel, thus confirming the previous work of Portevin, Pretet and Jolivet,[21] and of Chevenard.[22] Above the equi-cohesive temperature steel may, they state, be strengthened by the dissemination within the grain boundaries of a phase stable at the testing temperature.

Ranque and Henry also drew attention to the occurrence of intercrystalline breakdown at temperatures below the equi-cohesive temperature, which, for an austenitic steel of the 18/8 type, they place at approximately 1000 deg. Cent. Creep tests, particularly in the range 400 deg. to 500 deg. Cent., upon steels of this type often result in intercrystalline failure. This they explain as arising from the stability of gamma-iron of this composition at low temperatures being due to "inertia." Under the combined influence of stress and somewhat elevated temperature the gamma-iron is converted at the grain boundaries to another phase, which is probably in part martensitic. This transformation reduces the grain boundary strength below that of the crystals. This experience agrees with that of Krivobok,[23]

who states that the mechanical strength of nickel-chromium austenitic steels measured at high temperatures bears no relation to their behaviour under constant load, under which condition changes in structure are very pronounced and deep inter-crystalline penetration has been observed in alloys giving short life.

Dustin[4] remarks that in parts from service which have been examined in his laboratory intercrystalline fissures have always been associated with non-metallic inclusions. Creep tests which he has carried out have shown that a relatively small proportion of non-metallic inclusions in a specimen may reduce its life considerably.

McVetty[24] also has commented on the ageing characteristics of austenitic steels, and considers inadvisable their use under creep conditions. He includes in this condemnation all ageing alloys, with the exception of those age-hardening alloys whose hardness has been proved never to fall from the maximum which it attains. He argues that all other age-hardening alloys, if maintained sufficiently long even at comparatively low ageing temperatures, will, after attaining maximum strength, suffer a reduction in strength to a value little different from that prior to ageing. If a creep test is carried on for such a period that only the hardening stage is included in the test, then the measured creep rate being influenced by the hardening will be low, but creep will occur under service conditions, which would entail more prolonged exposure to stress, at a rate which will increase once the maximum strength has been passed. Again, in alloys which are hardened by precipitation the ductility may be reduced and the resulting brittleness militate against service applications.

Whilst being in complete agreement with the latter point, the present writer is unconvinced by McVetty's argument on the former point. In his argument McVetty makes the assumption that the "instantaneous creep rate" of an alloy during ageing is inversely as its Brinell hardness or tensile strength. Whilst there does not appear to be in the literature any direct evidence on this point, there is, as is pointed out elsewhere in this review, strong evidence that different structural considerations affect response to prolonged and short-time tests. Again, the investigations at the National Physical Laboratory upon materials for high-temperature service have shown that age hardening, as measured by the Brinell test at room temperature, occurs in certain complex nickel-chromium-iron alloys.[25] Nevertheless, the tempering of such an alloy for a sufficient period to develop its maximum hardness prior to a creep test was without appreciable effect upon its life under creep conditions at 800 deg. Cent., and certainly did not reduce its life in the manner which McVetty's remarks suggest. Therefore, the writer feels that unless McVetty is in possession of striking unpublished information on this point, he is unduly cautious in his attitude towards age-hardening alloys for creep service.

Returning for the moment to the "equi-cohesive temperature," one feels that attention should be drawn to the work of Singleton[26] on lead alloys, in which he attributes brittle intercrystalline failures in coarse-grained lead to the effect of unequal stress distribution, which he regards as inevitable in metal containing large grains of varying size. As coarse and irregular grain size is often met with in metals subjected to stress above the temperature of ready recrystallisation, it would appear not unlikely that an explanation of brittle fractures in metals in general might be arrived at which would not involve the amorphous-cement hypothesis.

As regards the measurement of creep resistance the situation still appears to be in the air. Whilst it is almost universally agreed that the fundamental method of testing the suitability of a material for a certain period of service (usually one or two million hours is required according to Dustin[4]) under a given stress at the service temperature is to subject that material to that temperature and stress for that period and ascertain whether it will endure without undue extension, such a course is generally regarded as prohibitively expensive. McVetty[6] is of the opinion that materials would be obsolete before the completion of the test, but that would surely depend upon the insight with which materials for test were chosen. The reason why some of the materials tested by Dickenson only twelve years ago have gone out of use is that these were materials already in service which were shown by the creep tests to be comparatively poor. At present the testing of materials under constant stress is being carried out in many laboratories, but in very few are tests carried on for the same period, or is the same quantity returned as the result of test. In other laboratories, again, tests are carried out under conditions of varying stress, approximating to a long-drawn-out tensile test, or under varying temperature. Of the papers to which we are giving special attention, that of Körber[3] gives the key to the present situation. He says, in effect, that if one admits the existence of a stress below which any creep which commences arrests itself in time, then one may experimentally determine at least an approximation to that limit by conclusions from comparatively short tests. If, on the other hand, one believes that creep at any stress will continue unendingly, then one must make very prolonged tests and design on "permissible creep rate." At the Kaiser Wilhelm Institut für Eisenforschung the existence of a definite creep limit is accepted, and, in consequence, short-time tests are almost invariably made. It is the opinion of the workers in that laboratory that a creep of less than 0·015 per cent. occurring between the third and sixth, fifth and tenth, or twenty-fifth and thirty-fifth hours of a test cannot lead to failure, except perhaps in materials suffering pronounced structural change on tempering at the testing temperature. Körber also attaches considerable weight to the 0·2 per cent. yield point, but only in respect of short exposure to temperature. He draws attention to the suggestion of Siebel and Ulrich[27] to measure the load causing 0·2 per cent. permanent deformation under conditions of stressing which would deform the metal at a rate of less than 0·0001 per cent. per hour.

Ranque and Henry[5] are also believers in short-time tests, but they favour the method proposed by Rohn[28] in which one effectively measures the temperature at which a specified rate of extension occurs under an applied constant load, the change in dimensions of the specimen controlling the temperature within the furnace. The apparatus, which is very fully described in the paper, allows of the detection of temperature changes of 1 deg. Cent. during thirty hours, this corresponding to an extension rate of 10^{-4} per cent. per hour, which is, they state, apparently the minimum creep rate considered by Pomp and Hoeger,[29] of the Kaiser Wilhelm Institut. Another modification of Rohn's method has been employed by Austin and Gier.[30]

Dustin,[4] whose contribution consists in the main

of a discussion of the relative merits of the diverse systems of testing employed in various laboratories, appears to belong to the school which believes in continuous creep at all stresses. He is definitely opposed to short-time tests and points out that :—

(1) Extremely variable results may be obtained from a tempered steel, such as a 0·5 per cent. Cr/0·5 per cent. Mo steel, in which the A_{1r} point occurs at about 500 deg. Cent., unless the material has been subjected to a prolonged treatment at 500 deg. Cent. before measurement of its creep rate. (Körber points out that in a 0·1 per cent. C steel the rate of extension under 5 kilos. per square millimetre at 500 deg. Cent. is reduced to a half by prior heating without stress for thirty-five hours.)

(2) Creep in the early stages is extremely susceptible to differences in prior thermal history, undetectable by other means.

(3) Experiments on opposed fibres and comparison of test results have shown that at least during the first twenty hours of test there are profound and irregular differences in the creep rate of specimens thought to be identical. These differences tend to become either zero or constant at longer periods of loading.

In opposition to "stage testing," he cites Eckhardt's[31] finding that a creep rate of 10^{-6} per cent. per hour between the fourth and eighth hours of a test may be produced by loads as different as 5 kilos. and 13·5 kilos. per square millimetre in the same steel at 450 deg. Cent. solely as a result of variations in the mode of application of the load.

He, therefore, concludes that tests lasting 100 to 150 hours on material in the normalised condition at suitable temperatures and loads which will give readily measurable extension, will be the most suitable for routine testing ; but when abnormal results arise in these tests he will employ tests of 1000 hours' duration. These will include both the N.P.L. and A.S.T.M. specifications. At the National Physical Laboratory,[32] it is considered satisfactory to assess the stress giving a creep rate of 10^{-5}in. per inch per day at the end of forty days, whilst the new American tentative specification[33] calls for measurement of two stresses, namely, those which give 1 and 0·1 per cent. extension in 1000 hours.

As has been pointed out above the main factors in deciding whether a short-time test is at all helpful are :

(1) Does a creep limit exist ? With reference to this point it should be stated that Moore and Alleman[34] found in tests on lead cable sheathing that graphs of stress against log. time taken to creep 1 per cent. show some flattening out with decrease in stress, thus indicating that there may be a creep limit, but they observed continuous flow under a load less than one-tenth of the ordinary tensile strength, thus demonstrating that such a stress limit, if it exists, is very low.

(2) Is the material structurally stable ? In addition to the occurrences of structural instability referred to above, reference should be made to McVetty's[6] experience in testing an alloy which, as a result of continued structural change involving decrease in volume, exhibited continuous negative creep over a period of 2500 hours.

(3) Is the resistance to load in the early stages of a creep test, or in a rapid stage test determined by the same structural considerations as is true creep resistance ? Alloys may certainly lie in a different order of merit as regards resistance to short and long-time tests, as is shown, for example, by Jenkins[35] in his study of the ball-hardness of metallic cadmium. Whereas cast cadmium is harder than rolled cadmium as measured by a prolonged test, in a short-time test it is softer. Again, Chadwick's[36] data for cold-rolled electrolytic zinc show that while the tensile strength is markedly increased by the early stages of rolling, the creep resistance is not, the tensile strength in this case corresponding to a rapid stage test. It has been shown repeatedly that coarsened steels are more resistant to creep, yet the effect of coarsening on the ordinary tensile properties is, if anything, adverse. The work of Krivobok[37] has shown that the amount of initial plastic deformation in a creep test, which is the main consideration in many short-time tests, is practically unrelated to the rate of creep at later stages of the test, in certain materials.

In this connection it would appear that the scheme of collaboration between the National Physical Laboratory and the Kaiser Wilhelm Institut[38] in comparing the results of short and long-time creep tests on the same steels, and the three-year programme of comparison tests proposed by the American Society of Testing Materials,[39] are ventures whose results should do much to clarify the present unsatisfactory position. Meanwhile, one cannot but agree with Gillett,[40] who, when faced with a conclusion[41] that short-time, up-step stage tests give the same result as true long-time creep tests, says : "This is another example of things that aren't so, broadly speaking. By chance they may coincide for a particular lot of steel, but there's plenty of definite evidence that generalisation from such an observation is quite erroneous. Such generalisations are often the result of wishful thinking. In time those who rely on the short cuts will find plenty of steels in which the alleged relation doesn't hold and will then back-water. In the meantime, plenty of harm is done by spreading a false gospel to engineers."

With regard to the interpretation of long-time tests, Kanter[42] has drawn attention to the fact that creep rates are often stated without reference to the initial plastic deformation, and objects on the score that the latter may sometimes exceed permissible deformation in service. The method of expressing creep data put forward by McVetty appears to meet this point. McVetty,[6] for the purposes of comparison and extrapolation, substitutes for the measured creep curve a straight line whose equation includes terms for the initial deformation and the measured creep rate. He then plots these values against stress and computes from the interpolated values a series of curves giving stress against total deformation for each of a series of lives, from which a designer can read the stress which will allow total deformation to be kept within a fixed limit for the desired service life. In his paper McVetty treats results of Tapsell and Johnson[43] for forged and cast steels in this manner. The writer notes that these final charts have been extrapolated to periods up to twenty years, and cover a range of plastic deformation up to 1 per cent. Whilst this appears a quite sound course for these steels at the temperatures involved, one wonders to what extent it is justifiable to assume a value for total elongation to fracture from an uncompleted creep test at low stress, as it is known from creep tests at higher stresses which have been carried to completion that the supportable elongation tends to decrease with fall in applied stress. It appears quite possible, to look at this point from

another angle, that two different materials may have the same creep rate and yet break at very different elongations. If this is so, then designer's formulæ which take into account either creep rate alone, or initial deformation and creep rate together, without considering attainable elongation, may sometimes prove to be dangerously misleading.

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Sulphur Prints and Sulphide Inclusions.

By L. NORTHCOTT, Ph.D., M.Sc., F.I.C.*

THE distribution of sulphur in a steel section may be determined by means of a sulphur print. The brown dots in such prints are generally conside ed to be representative of particles of the sulphides of manganese and iron occurring as insoluble impurities in the steel. The opinion has, however, been expressed that a preponderating amount of the sulphur is in actual solution in the steel, probably as manganese sulphide.

With a view to obtaining some information on the matter, an examination was made of a polished steel surface and a sulphur print obtained from it. The steel used was a rimming steel ingot section of the following composition :—Carbon 0·09, silicon 0·005, manganese 0·365, sulphur 0·061, phosphorus 0·019 per cent. The cavities in the rimming steel served to locate the areas to be examined, and were associated with local sulphur segregations. Three areas, typical of low, medium, and high sulphur content, were selected. Each area was photographed at a magnification of ×25, and a photograph (reversed) at the same magnification was obtained of the corresponding area in the sulphur print. The three pairs of photographs are shown in Figs. 1 to 6.

Correlation of sulphur print with the steel is made somewhat difficult by the diverse effects obtained on the sulphur print by what appear to be similar inclusions in the steel. This variation may be due to poor contact between paper and steel, uneven acid content in the paper, or inclusions having different sulphur contents. The sulphur print marks were light brown or dark brown, and appear grey or black in the photographs. Preliminary etching to remove any possible surface film on the steel did not affect this. Tests made with paper containing different quantities of the standard acid solution showed that the proportion of pale brown areas in the sulphur print increased with the degree of "dryness" of the paper. When the paper, after being soaked in acid, was drained, but not dried between absorbent paper, no pronounced difference in colour occurred.

MICROSTRUCTURE OF THE INCLUSIONS.

Micro-examination of the steel offered some clue to the large difference in size of the sulphur print marks. There were found to be at least five types of inclusions, namely, silicate streaks and globules; dove-grey sulphide globules; a brownish-yellow intergranular network of iron sulphide which occurred mostly near the cavities, Fig. 7; a eutectic-like structure the globules of which were similar to the normally occurring dove-grey sulphide deposits, Fig. 8; and large globules showing a duplex structure possibly consisting of mixed sulphide solutions, Fig. 9. The silicate streaks, except when associated with ferrous sulphide, appeared not to affect the photographic paper. Of the sulphides, the iron sulphide network was the most effective in producing dark marks on the paper. The eutectic-like deposits were similar to those observed by Arnold and Bolsover† in small experimental ingots containing aluminium; similar ingots without aluminium did not show these structures. The ingot examined in the present work was, however, free from aluminium. Similar structures have recently been described by Vogel and Baur‡ as droplets of MnS . FeS complex separating in the liquid state, giving the appearance of a eutectic by reason of their mode of distribution. In the large sulphide globules showing a duplex structure, the dove-grey and pale brown constituents occurred either in massive form or as in a eutectic, both types being seen in Fig. 9. It is possible that the structures result from the co-existence of the α and β solid solutions of the FeS–MnS system, the eutectic of which melts at 1164 deg.

* Communication from the Research Department, Woolwich.

† *Journal* of the Iron and Steel Institute, 1914, No. 1, page 396.

‡ Archiv fur das Eisenhuttenwesen, 1933, 6, 495.

SULPHUR PRINTS AND SULPHIDE INCLUSIONS

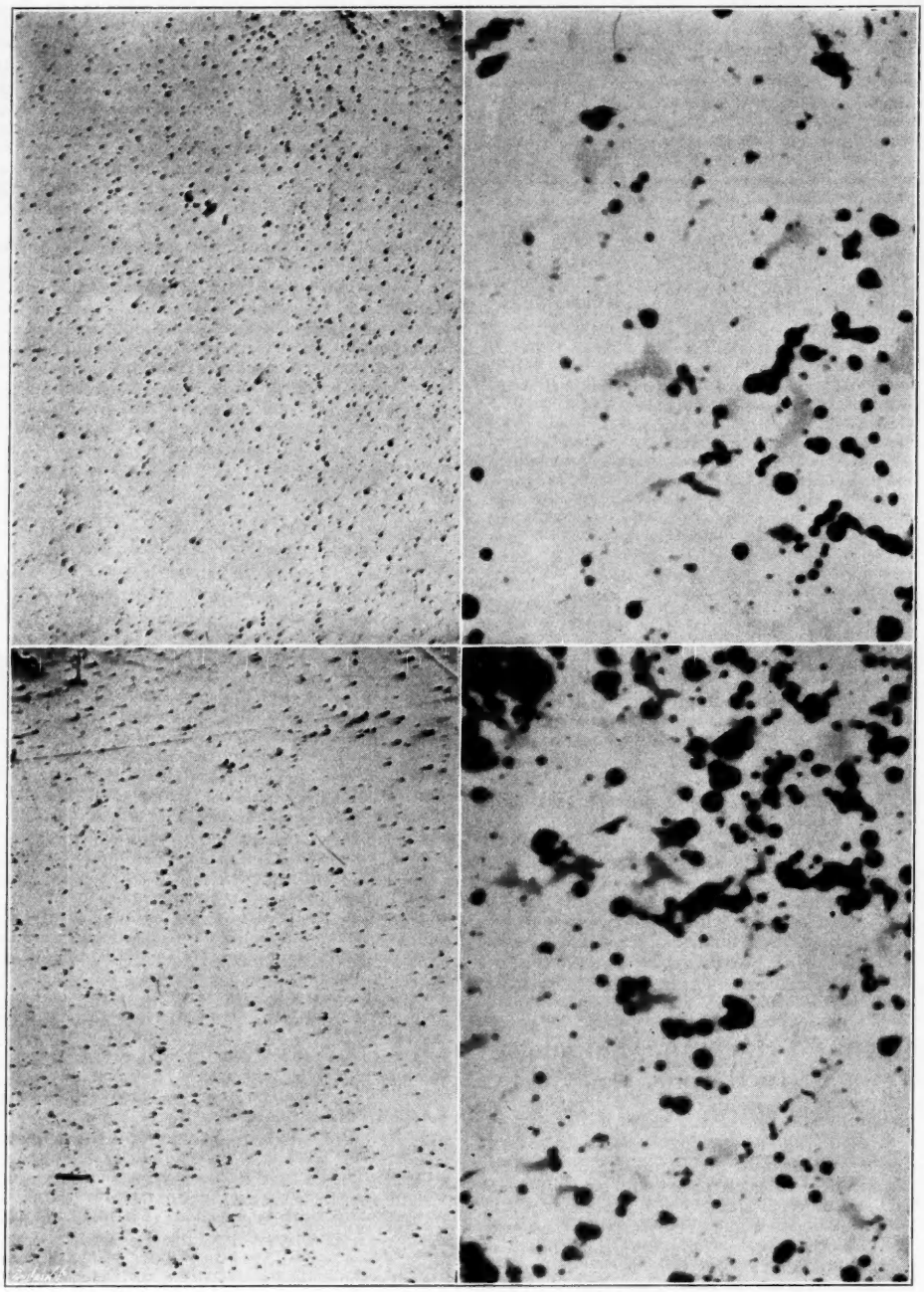


FIG. 1 (ABOVE)—STEEL (1) ×25
FIG. 3 (BELOW)—STEEL (2) ×25

FIG. 2 (ABOVE)—SULPHUR PRINT (1) ×25
FIG. 4 (BELOW)—SULPHUR PRINT (2) ×25

Cent. according to Shibata.§ The appearance of such areas was different from that of slag inclusions. It is probable that the normal dove-grey sulphide inclusions also consist of a solid solution of iron sulphide and manganese sulphide, since the globular form in which they occur is not typical of material melting at such a high temperature as does manganese sulphide, namely, 1620 deg. Cent. It is remarkable that sulphur occurs in such different forms not only in one ingot, but in a section only a few inches in length.

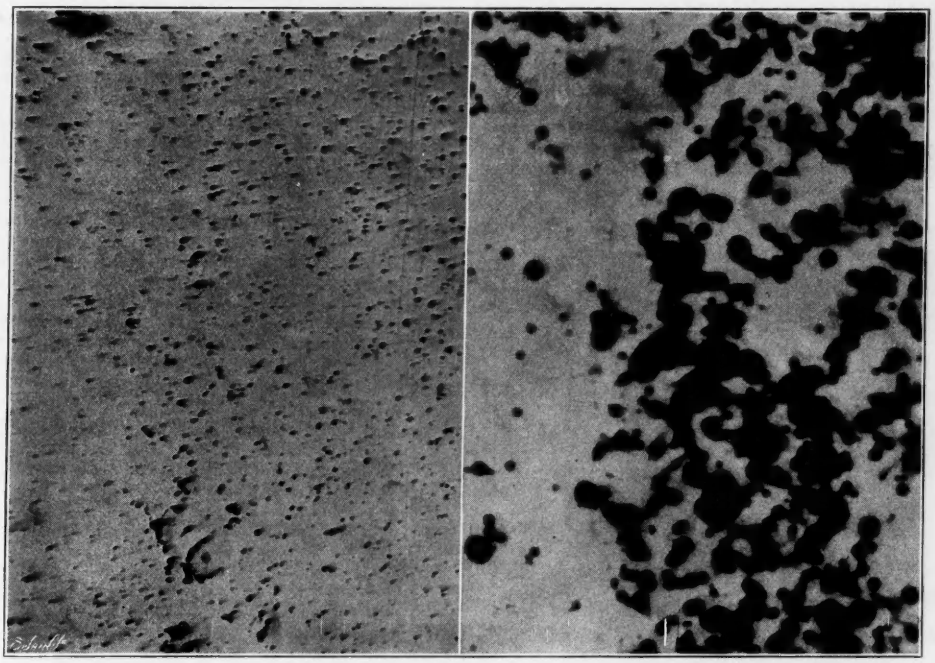


FIG. 5—STEEL (3) ×25

FIG. 6—SULPHUR PRINT (3) ×25

Direct comparison between sulphur print and steel section may be facilitated by sulphur printing on photographic film instead of on paper. A clear print may be obtained and the finished film can be placed over the steel section to facilitate the identification of the marks. An interesting modification of

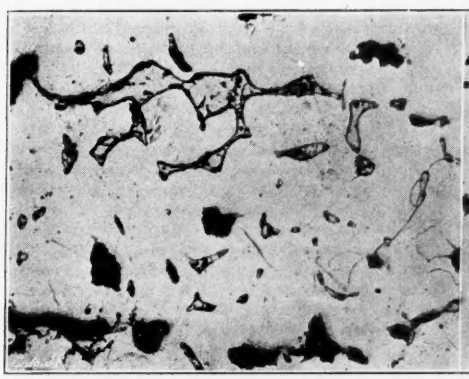


FIG. 7—INTERGRANULAR IRON SULPHIDE ×100

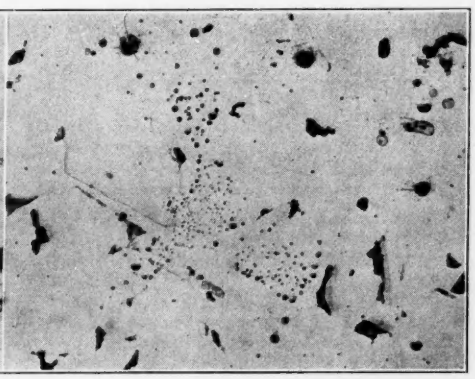


FIG. 8—EUTECTIC TYPE FORMATION OF SULPHIDES ×100

§ THE METALLURGIST, 1929, 5, 21.

this method which has been developed in Germany was recently described in THE METALLURGIST.‖

CONCLUSIONS.

From a study of the steel sections and sulphur prints it was observed that the sulphur prints show a number of dark spots, considerably larger yet fewer in number than the inclusions in the steel, and there was no constant relation between them in size. The sample showing local segregation of sulphide inclusions yielded a print in which the segregation was appreciably exaggerated. It was found possible to identify a large number of the sulphide areas on the sulphur prints with inclusions in the steel. Iron sulphide inclusions were most effective in marking the paper. Of other sulphide

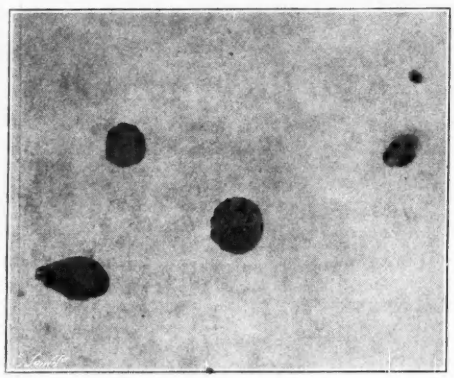


FIG. 9—DUPLEX STRUCTURE OF MIXED SULPHIDES IN LARGE GLOBULES ×500

inclusions those of duplex structure containing iron sulphide, or solutions rich in iron sulphide, were most active, while silicate inclusions, of course, produced no effect.

The intensity of the sulphur print thus depends on the composition of the inclusions (manganese sulphide being less effective than iron sulphide) as well as on the conditions (such as the acid content of the paper) under which the print is made.

Progress in the Sintering of Metal Powders.

W. D. JONES, M.Eng., Ph.D. (Liverpool).

AN article in THE METALLURGIST, February 24th, 1933, was devoted to the sintering of metal powders. Considerable progress has been made in this branch of metallurgy since the date of the article, and therefore a reconsideration of the subject may be of value. As an example, in that review stress was laid on the fact that alloys manufactured by sintering powders are generally porous. At the present time, however, this cannot be considered necessarily true, any more than is the belief of two years ago that powder metallurgy is confined to the processing of tungsten products and certain other materials of high melting points. Many of the former difficulties and limitations of the process have been overcome, and when it is stated that the present consumption of powdered metal for the manufacture of sintered products amounts to not less than 100 tons per month, it will be realised that the process deserves more attention than it appears to be receiving.

‖ THE METALLURGIST, 1933, 9, 43.

The manufacture of metal powders is not a new industry. An outlet for these materials for many years has been found in the form of gilding or silvering solutions and varnishes. More recently aluminium, copper, bronze, tin, nickel, and zinc have been employed in this manner as a wash or protective coating sprayed on, either suspended in a liquid medium or solid at elevated temperatures. The chemical industries are a considerable consumer of metal powders, and recently the physical and radio worlds have turned their attention in this direction in the manufacture of low and high-frequency transformer cores. This article, however, deals only with the entirely new application of powders for the production of solid metallic shapes by sintering.

The sources of information on the subject are mainly German and American, and much of the following has been extracted from the following publications:—

(i) *Metallkeramik*, F. Skaupy, 1930, Berlin.
(ii) Schlecht, Schubardt, and Duftschmid, *Z. f. Elektrochemie*, 1931, 37, 485–491.
(iii) Duftschmid, Schlecht, and Schubardt, *Stahl u. Eisen*, 1932, 35, 845–849.
(iv) Sauerwald and St. Kubik, *Z. f. Elektrochemie*, 1932, 38, 33.
(v) W. D. Jones, *Journal* of the Iron and Steel Institute, 1934, II, 429–437.
(vi) Charles Hardy, *Metal Progress*, 1932, 22 (1), 32–37, 80.
(vii) Charles Hardy, *Engineering and Mining Journal*, 1933, 134 (9), 373–374.

The following are the main points for consideration in the sintering of metal powders.

I. METHODS OF POWDER MANUFACTURE.

(*a*) Reduction of an oxide or decomposition of a metallic salt is conducted practically at elevated temperatures in a reducing atmosphere. Commercially, the products are seldom pure; the author has examined several samples of reduced iron which were most unsatisfactory in this respect. Considerable control over particle size can be secured mainly by temperature control, but at the same time it appears to be difficult and inefficient to produce very fine powders which are reasonably free from oxide.

(*b*) Precipitation from solutions (*e.g.*, tin from the tetrachloride) enables close control over particle size and oxide content to be secured, but, again, this is a process which is hardly economic for large outputs. Titanium powder has recently been prepared by deposition from solutions of the tetrachloride in xylene by the addition of sodium.

(*c*) Metals with low melting points, as, for example, tin and lead, are extensively prepared in the form of powders by atomising the molten metal with compressed air. Superficial oxidation is impossible to prevent unless a neutral gas is employed in the place of compressed air, and the resultant powders find applications therefore mainly in directions where such oxidation is unimportant.

(*d*) Preparation by crushing probably accounts for more ranges of powder than any other. The shape of the powder grains depends greatly on the nature of the solid material to be crushed. Malleable metals,

such as copper, bronze, and aluminium, produce flakes rather than powders. Such powders, with an axis ratio of anything from 5 to 25 are hardly suitable for sintering. For this purpose it has been found more convenient to use as raw materials the metals in a hard and brittle condition. It will be realised that no material could be more suitable in this connection than an electro-deposit. By suitably controlling the conditions of deposition (see D. J. Macnaughtan, *Journal* of the Iron and Steel Institute, 1924, I, 409), it is possible to obtain metallic deposits saturated with hydrogen which are particularly brittle. As an impurity, the hydrogen is of no importance in this particular connection, and is removed during sintering. A less expensive and more adaptable method is that of producing, by manipulating the conditions during electrolysis, a spongy rather than a brittle deposit. Such a deposit, after washing, can be reduced to a fine powder by only a slight amount of mechanical rubbing or grinding. The following metal powders have been produced by electrolytic methods, and can be obtained commercially :—Copper, tin, zinc, nickel, iron, chromium, molybdenum, tungsten, and silver. Powders made by crushing brittle or spongy deposits generally consist of rough sharp particles, angular in shape.

(*e*) Iron powder is manufactured in large quantities by the decomposition of the carbonyl. The particles are smooth and globular, and can be produced in almost any degree of fineness likely to be commercially necessary. This powder finds considerable application in the manufacture of iron alloys, but probably its most extensive use is as a magnetic material in the electrical and radio industries. A micro-section of carbonyl-iron shows that the particles are considerably cored and apparently far from homogeneous.

II. Shape, Size, and Density.

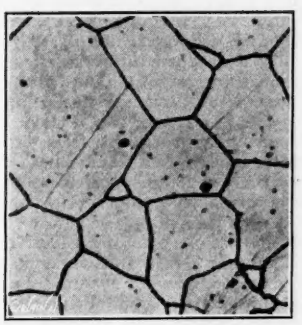
For the purposes of high cohesion it is probably desirable to employ irregular angular shaped particles, since they might be expected to interlock well. From the points of view of easy packing and compression, however, globular particles would be better. The most satisfactory results are probably obtained by a compromise between these two shapes, a fact which may explain the successful use of spongy electro-deposits. There is room for investigation in this field.

The shape of the particles controls the packing density. For economic reasons, this should be kept as high as possible. As an example, suppose it was desired to compress two copper powders, of apparent densities 2·75 and 1·0 respectively, to briquettes which have a density of 8·25. For a briquette 2in. high the filling heights necessary would be 6in. and 16·5in. respectively. In general, the coarser the powder the greater is the packing density, but the two figures given for copper can be regarded as typical. The amount of air entrapped during compression increases as the packing density decreases. This is a sound reason for not going too far in the direction of fineness.

Powders are graded by screens, bolting cloths, and liquid or gaseous elutriation. (For detailed particulars of gaseous elutriation, see Skaupy, *Metallkeramik.*) Commercially, powders conforming to specified screen analyses between 100 and 500 mesh can be purchased. The rate of alloying together of two metal powders is governed more by the specific surface than by the mesh. Recent investigations of methods for calculating the specific surface of powders are to be found in a paper by H. Heywood, Inst. Mech. Engineers, 1933, 125 (2), 383–459.

III. Purity.

Apart from oxygen the chief impurities in metal powders are those incidental to the manufacture. For example, in the production of copper powder by crushing of electro-deposits very pure copper should be available. Nevertheless, a certain amount of iron and possibly silicon will have originated from the crushing equipment. In the case of iron powder from electro-deposits, commercial samples have been obtained very low in oxygen and carbon, but containing as much as 0·08 per cent. phosphorus, an effect due probably to the transfer of phosphorus from the anodes. There is no reason why a purity of less than 99·5 per cent. should be obtained on any commercial crushed metal powder and frequently much better figures are encountered. In iron powders produced by other processes carbon is often a serious impurity. Although it is claimed to be possible to produce carbon-less carbonyl-iron, the samples which the writer has examined have invariably contained more than 0·15 per cent. An approximate value for the oxygen content of iron powder can very quickly be obtained by solution in dilute nitric acid (1+3). The amount of residue of iron oxide, while not representing necessarily the total oxygen content, forms a ready means of comparison with materials of known oxygen content. The majority of iron alloys which the writer has prepared without deoxidation of any kind from electrolytic iron powder have, after sintering and sectioning for the microscope, displayed iron oxide inclusions to an extent of three to four times the amount usually found in Armco iron. The annexed figure shows the microstructure of an alloy of iron



IRON-TIN ALLOY SINTERED AT 1200 DEG. CENT.

with 0·2 per cent. tin, sintered without deoxidation at 1200 deg. Cent., from 500-mesh iron and 200-mesh tin compressed at 35 tons per square inch. The magnification is 500 diameters.

IV. Compression.

The density and strength of the sintered material is largely controlled by the amount of compression prior to sintering. Commercially, this varies between 1 ton and 100 tons per square inch. The maximum pressure available is a serious limitation to the production of pressed shapes. Quite apart from the extremely high forces involved as soon as sizes become large, there are difficulties due to the fact that, unlike the case of a liquid, the pressure is not equally distri-

buted throughout the bulk, resulting frequently in shapes with loosely compressed cores or bases. For this reason a cylinder of iron powder compressed from the top will, after sintering, often take the shape of an inverted cone. If the form to be pressed is designed with a view to overcoming these difficulties it appears to be commercially possible to manufacture parts as large as 6in. diameter and 6in. deep. Better results are obtained if the sintered articles are recompressed to close up pores and then re-annealed for quite short periods. Densities equal to and sometimes greater than those of the cast material can then be obtained. There is scope for inquiry into methods of simultaneous sintering and compressing of powders under definite known pressures. By this means maximum pressures would be considerably reduced and possibly a better control over the grain growth would be secured. The work of Sauerwald and St. Kubik indicates the importance of grain size control. Working with copper and iron powders which were compressed over a range from approximately ½ ton to 2 tons per square inch for thirty seconds and annealed at various temperatures in hydrogen for thirty minutes, measurements were made upon density, hardness, strength, grain size, and electrical conductivity. In each case maximum density and strength was obtained at moderate temperatures (600 deg. to 900 deg. Cent.), and in the case of copper powder the best results were obtained by using only moderate pressures; an illustration of the interdependence of grain growth, temperature, and lattice deformation.

In contrast to these conclusions it is interesting to consider the work of Schlecht, Schubardt, and Duftschmid. Carbonyl-iron powder in bulk without any compression was heated in an atmosphere of hydrogen for twenty-four hours at temperatures from 400 deg. to 1200 deg. Cent. Sections examined under the microscope showed a gradual agglomeration of particles from 400 deg. Cent. (density 3·2) to 700 deg. Cent. (density 5·6) when the material became continuous although very porous. The density attained a temporary maximum between 800 deg. and 900 deg. Cent. (6·1 at 800 deg. Cent.) and then, due to the volume change at the A_3 transition, diminished again to 5·3 at 910 deg. Cent. with an increase in porosity. At 1200 deg. Cent. the density became 6·3 and the metal was only slightly porous. After this material had been worked the density was raised to 7·8 and under the microscope was indistinguishable from metal that had been cast. Thus these workers claim that preliminary compression is unnecessary in order to secure sound billets. Granted that this is so, it is difficult to understand how a commercial process based on this method would be successful, since the large shrinkage during sintering would not be simple to control exactly in different directions, and the size of the articles before sintering would necessarily involve large furnaces.

V. SINTERING.

Temperatures and time of sintering depend almost entirely on the nature of the material. For pure materials a time of two or three hours at a temperature two-thirds of the melting point of the metal is a rough guide. For carbonyl-iron a temperature of 1200 deg. to 1300 deg. Cent. for two to three hours is cited. In the case of alloys additional time may have to be allowed for interdiffusion of the constituents. If one of the components is above its melting point during sintering then two or three hours may again be sufficient, provided the powders were thoroughly mixed. Should both metals be solid during sintering then twelve to forty-eight hours may be necessary before the product is completely homogeneous.

Generally it is necessary to exclude air during the preliminary stages of sintering, while the best results are obtained by heating in a reducing atmosphere. Probably good results could also be obtained by mixing with the powders a deoxidant evolving a gaseous product. In the case of carbonyl-iron, owing to the carbon content of the material, such circumstances occur automatically. Carbonyl-iron employed commercially for sintering frequently analyses 1 to 2 per cent. oxygen and similar amounts of carbon, either in the elemental form or as surface adsorbed carbon monoxide. On heating, water and methane are evolved at 200 deg. to 300 deg. Cent. and at about 450 deg. Cent. the reaction between carbon and oxygen commences. At low temperatures carbon dioxide is the main product of the reaction; the relative amounts of carbon dioxide and carbon monoxide at different temperatures are of the following orders:—500 deg. to 600 deg. Cent., 66 per cent. CO_2, 34 per cent. CO; 700 deg. to 800 deg. Cent., 35 per cent. CO_2, 65 per cent. CO; 900 deg. to 950 deg. Cent., 10 per cent. CO_2, 90 per cent. CO. In the case of small articles at 1000 deg. Cent. the evolution of gas is complete in about half-an-hour. It will be seen that in this particular case the material during the initial important stages of sintering is enveloped in a protective reducing atmosphere, thus making an external reducing atmosphere hardly necessary. In the case of carbonyl-iron it is claimed that it is possible so to adjust the carbon content that the product is completely deoxidised and decarbonised. While this is undoubtedly possible, it would necessitate in industry a rather close control over powder manufacture, compression, and rate of rise of temperature.

As an example of results obtained industrially the following can be cited:—

(i) (Duftschmid, Schubardt, and Schlecht). 170 kilos. of carbonyl-iron containing 1·26 per cent. C and 1·36 per cent O_2 were compressed to a density of 3·3. The pressing, 800 mm. high and 280 mm. diameter, was sintered for four hours at 1100 deg. Cent. in a gas-fired muffle. After sintering the density was 6·9, the height 624 mm., and the diameter 224 mm. The block was then forged under a steam hammer into a cube 120 mm. by 120 mm. without the slightest signs of cracking. The average carbon content was 0·37 per cent.

(ii) Sheets of carbonyl-iron produced by sintering and hot rolling, 1·0 mm. thick, had an Erichsen value of 12·25 mm. Normal deep-draw mild steel strip varies (according to Duftschmid, Schlecht, and Schubardt) between 10·0 mm. and 11·0 mm. for the same thickness.

(iii) (Charles Hardy). Bronze wires have been drawn from a sintered ingot with a tensile strength 30 to 40 per cent. greater than that of wires of like composition made from cast wire bars.

(iv) (W. D. Jones). Alloys of iron or other metals containing several percentages of volatile metals (zinc, antimony, and arsenic) have been made with ease. This would be difficult, uneconomic, or impossible by casting methods.

It is realised that powdered metals in a pure form are expensive and that generally their use is limited to cases where the expense is offset by some specific

and peculiar advantage. Some of these advantages are as follows (Charles Hardy) :—

The porosity of the product is under control. Bronze bearings containing graphite with crushing strengths up to 30 tons per square inch can be manufactured which will adsorb up to 30 or 40 per cent. of oil by volume. Silver or stainless steel table ware can be produced with half the density of that made from strip. Metal of the highest purity can be obtained. Alloys of constant composition are produced. The electrical conductivity of small parts for the telephone industry, for example, can be maintained with a greater uniformity than by any other method. Alloys and mixtures can be produced of metals immiscible in the liquid state or having widely differing melting points. Mixtures of metal and non-metal can be effected. Bimetal parts or strip are easily made. The size of the completed article can be held to very close limits ; it is possible to compete in this connection with the very best die castings. Manufacturing losses are very small, since there are none corresponding to the operations of slagging, cropping, oxidation, or trimming.

Locomotive Tires.

IN a paper which is published in the November and December issues of *Metals and Alloys*, L. E. Grant discusses observations which he has made over a number of years on locomotive tires with the object of ascertaining the principal causes of failure. The observations relate to American practice, and a large number of tests have been made, both on tires which have failed and on tires which have worn out in service.

In general there are four types of tire failure, namely, the snap break, progressive fracture, shelling, and failures due to defects in the steel.

Snap Breaks.—In snap breaks there is no indication of progressive fracture or defect of any kind, which suggests that the metal is unduly brittle or that the stresses are abnormal, or possibly both. They may be caused by stresses induced by too great a shrinkage, the customary allowance for which should result in a stress of approximately 30,000 lb. per square inch. The final load can be influenced by many factors, such as variations in the roundness of both tire and wheel, taper in the bore, roughness of the surfaces, and accuracy of the dimensions. Snap failures may also be caused by improperly counterbalanced wheels, rough tracks, flat spots, wheels out of round, or any other condition that sets up a severe hammering on the tire, but most of these causes can be eliminated. As snap breaks, like many other failures, generally occur in tires almost worn out, the thickness at which a tire is condemned has an important bearing on the prevention of this type of failure.

Progressive Fractures.—Progressive fracture failures may start either in the bore or the tread. In the former case they usually start at or near one side, and in the latter always in the heat checks in the flange section. The outside edge and the centre of the tread may show heat checks, but fractures do not generally occur there. The flange is subjected to severe braking effects, and the frictional heat produces checks from which the fractures originate. In flange failures the smooth surface of progressive fractures is rarely seen. Because of the high internal stress, the spread of a fracture in tires is more rapid than in other locomotive forgings, the area of progressive fracture averaging about 5 and 50 per cent. respectively of the cross-sectional area in the two cases. Progressive fractures beginning in the bore appear to start chiefly on the flange side of the tire, and it has been suggested that this may be due to the sharp edge of the wheel acting as a stress raiser, some support being lent to this view by the observation that when heat-treated tires are used, it is customary to round off the edges of both tire and wheel centre, which tends to prevent such failures. Service conditions resulting in severe impacts are also contributory causes of failure. Small tears adjacent to the origin of the fracture are often revealed by deep etching, but it is not known whether they develop from tool marks or are the result of stress. Some bore failures start in flat spots, at the edge of the tire on either the flange or the outside edge, which have been caused in driving the tire on or by a blow in service. Fractures in the centre of the bore which occasionally occur have been associated with a small defect, such as a crack where a chip has broken off. A concavity in the bore due to machining, or which has developed in service, leads to extra stress occurring in that position when the tire is worn and is conducive to failure.

Shelling.—More common than either of the failures so far referred to is shelling, which takes the form of a surface breakdown of the metal, usually in the centre of the tread, sometimes spreading to one side, but never affecting the flange itself. The defect sometimes spreads a considerable depth into the metal, but does not often lead to complete failure, because the tire is removed before the disintegration has proceeded too far. It has been attributed to defective metal, although the nature of the defect has not been defined. It has also been attributed to service conditions, but if this were wholly correct it is difficult to account for one tire on an engine shelling while the others remain intact. There seems to be evidence to show that service conditions, especially in cold weather, can cause shelling, and in this connection heat-treated tires have been reported as a remedy. Abnormal metal conditions are frequently revealed by microscopic examination of shelled areas, and a number of examples are illustrated. The conclusion is reached that whilst defective material cannot be disregarded as a possible cause of this form of trouble, it is not the sole cause, and much further work is required before a full explanation is possible.

Steel Defects.—The defects which may occur in the steel from which the tires are made do not differ from similar defects which may be encountered in other forgings. Internal cracks may be due to secondary pipes, for they are commonly associated with considerable segregation. Other defects originate from blow-holes and internal seams. Such defects account for a moderately large percentage of the total failures, the proportion being larger than it is from similar defects in other locomotive forgings. This is not attributed to the production of more defective ingots, but to the method of tire manufacture not being well adapted to the detection of defective ingots.

In America, locomotive tires are generally purchased to the A.R.A. specifications, covering the three grades indicated in the accompanying table. Most of the steel is made by the open-hearth process, and while it may be acid or basic, the former is more usual. One of the two chief methods by which tires are produced consists in flattening an ingot which is large enough for one tire only, punching out the centre, hammering partly to shape, and finishing by rolling. The work is carried out above the critical

temperature range, and the tires are cooled in air. In the other method slices of appropriate size are cut from a specially shaped bottom-poured ingot, and, after heating, are flattened and punched as in the first method. Without any hammering, however, they are transferred to the rolling mill for finishing,

Grade	A	B	C
Service	Passenger	Freight	Switch engines
Composition, per cent.—			
Carbon	0·50–0·70	0·60–0·80	0·70–0·85
Manganese	0·50–0·75	0·50–0·75	0·50–0·75
Phosphorus	0·05 max.	0·05 max.	0·05 max.
Sulphur	0·05 max.	0·05 max.	0·05 max.
Silicon	0·15–0·35	0·15–0·35	0·15–0·35
Mechanical properties—			
Tensile strength, lb. per square inch	105,000 min.	115,000 min.	125,000 min.
Elongation, per cent.	12	10	8
Reduction in area, per cent.	16	14	12

and cooled in air. Tires made in this way possess a coarser dendritic structure than those made by the first method.

It is the practice to carry out tensile tests on bars forged to about 1¼in. square from small test ingots, approximately 4in. square and weighing about 90 lb., cast at the same time as the metal from which the tires are made. Although the test bars are reduced to about the same extent as the tires, the properties obtained in them vary considerably from those obtained in test pieces cut from tires. Tests on specimens cut from tires afford evidence that with steels of the highest quality there is less difference in the properties of bars cut parallel and normal to the direction of rolling than there is in steels of lower grade. Tires which have been worn out in service have been examined with the object of ascertaining in what respects they differed from tires which had failed, and one of the principal differences noted was that free ferrite was not as noticeable in the former. The majority of failures were in class A tires, in which the ferrite is most predominant, which suggests that this constituent has a weakening effect. Out of forty-seven failed tires, twelve failures were attributed to defects in the steel, seven were snap breaks, eleven were progressive fractures originating in a heat-checked flange, fourteen were progressive fractures beginning in the bore (eleven on the flange side and three near the centre), and the remaining three were due to miscellaneous causes under other headings. As no accurate record was available, failures due to shelling were not included. Disregarding defective tires and two miscellaneous failures, only six of the remaining thirty-three had carbon contents in excess of 0·70 per cent., and only three of these contained more than 0·72 per cent. carbon. Of the remainder, two contained less than 0·60 per cent. carbon, while the other twenty-five were in the range 0·60 to 0·70 per cent., and it was concluded that tires for both freight and passenger engines should have a carbon content in the range of approximately 0·70 to 0·80 per cent.

Alloys of Low Thermal Expansibility.

THE use of alloys of low thermal expansibility for standards of length, bimetallic strip and other special purposes originated from an investigation made in 1896 by Guillaume, undertaken as the result of the discovery that the expansion of a bar of 30 per cent. nickel steel was about one-third less than that of platinum, which had a coefficient of only 9×10^{-6}. Previously, Benoit had found that a steel containing 22 per cent. nickel and 3 per cent. chromium had a coefficient as large as 18×10^{-6} or about that of bronze. Guillaume's study produced an iron alloy containing 36 per cent. nickel and small amounts of manganese, silicon, and chromium (total about 1 per cent.), which became known as "Invar." This alloy had a coefficient of expansion of $1\cdot2\times10^{-6}$. Improvements have since been made by means of control of thermal and mechanical treatment to ensure stability of the alloy. A later development was the production of "Elinvar," a 35 per cent. nickel-iron alloy containing about 10 per cent. chromium and 2 per cent. tungsten (coefficient of expansion about $6\cdot0\times10^{-6}$), which has been of considerable value as a material for hair springs in horological instruments, since its modulus of elasticity is practically independent of temperature between −50 deg. Cent. and 100 deg. Cent. The need for materials of particular expansibilities for special uses led to the finding of alloys such as "Platinite" (nickel with 46 per cent. chromium and 0·15 per cent. carbon) for wires to be sealed in glass.

The subject is an important one in view of the widening application of the principle of differential thermal expansion in the electrical industry and of the apparent fundamental connections between expansibility, magnetic properties, and thermal transformations of the alloys. During recent years, Masumoto, in Japan, has carried out the examination of a large number of alloys, and in 1929 obtained one having a smaller coefficient of expansion than that of fused silica. This contained 63·5 per cent. iron, 31·5 per cent. nickel, and 5 per cent. cobalt, and was named "Super Invar" (see Sci. Rep. Tôhoku Imp. Univ., 1931, 20, 101). Masumoto proposed a theory of small expansibility, which holds that the property in a ferro-magnetic alloy is correlated with the ratio of the saturation magnetisation I_o to the critical temperature θ, the greater the ratio the smaller the coefficient of expansion.

In the iron-cobalt system no alloy having a very small expansibility exists, but on the basis of the above hypothesis, and from the magnetic properties of the gamma phase of the system, it would be expected that the addition of a third element capable of lowering the critical point would markedly decrease the expansibility.

The results of some later work by Masumoto on these lines have now been published,* dealing with the study of the ternary alloys of cobalt, iron, and chromium. Cylindrical bars were used, cast from alloys prepared by melting the pure metals in alumina under a hydrogen atmosphere, the tests being made with the Honda dilatometer (Sci. Rep., 1924, 13, 101).

The detailed results from 105 alloys are shown graphically in Fig. 1. The curves refer to the mean coefficient of expansion between 20 deg. and 60 deg. Cent., determined from the heating curve. A ternary space diagram was also constructed and a projection of the curves of equal expansion coefficient on the basal plane of the model is given in Fig. 3. These show that with addition of chromium to the cobalt-iron alloys, the coefficient of expansion diminishes, first gradually and then rapidly, to an abrupt minimum at 9 per cent. chromium, and then again rapidly

* Hakar Masumoto, "On the Thermal Expansion of Alloys of Cobalt, Iron, and Chromium," Sci. Rep., Tôhoku Imp. Univ., 1934, 23, 265–80.

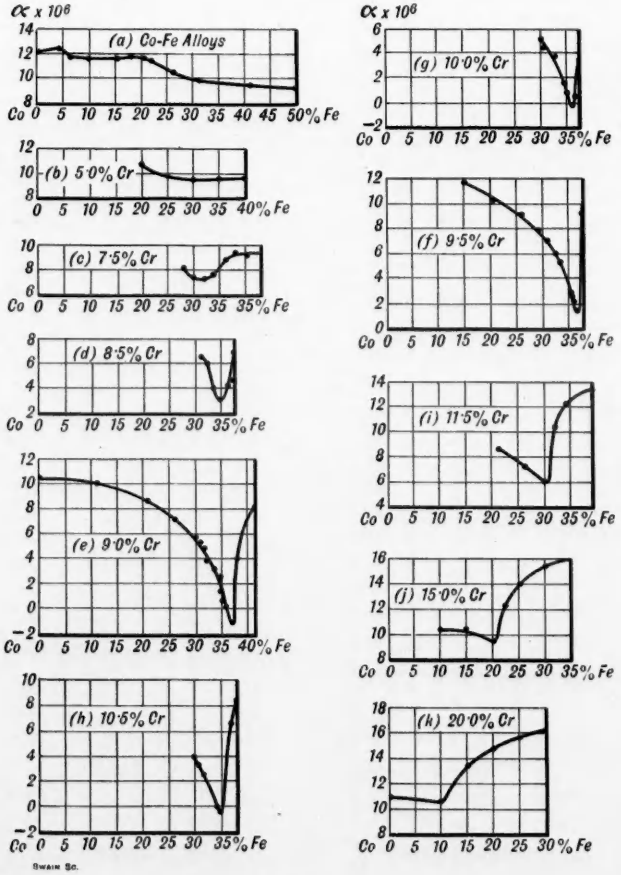


FIG. 1—*Thermal Expansion. Effect of Addition of Chromium to Cobalt-Iron Alloys.*

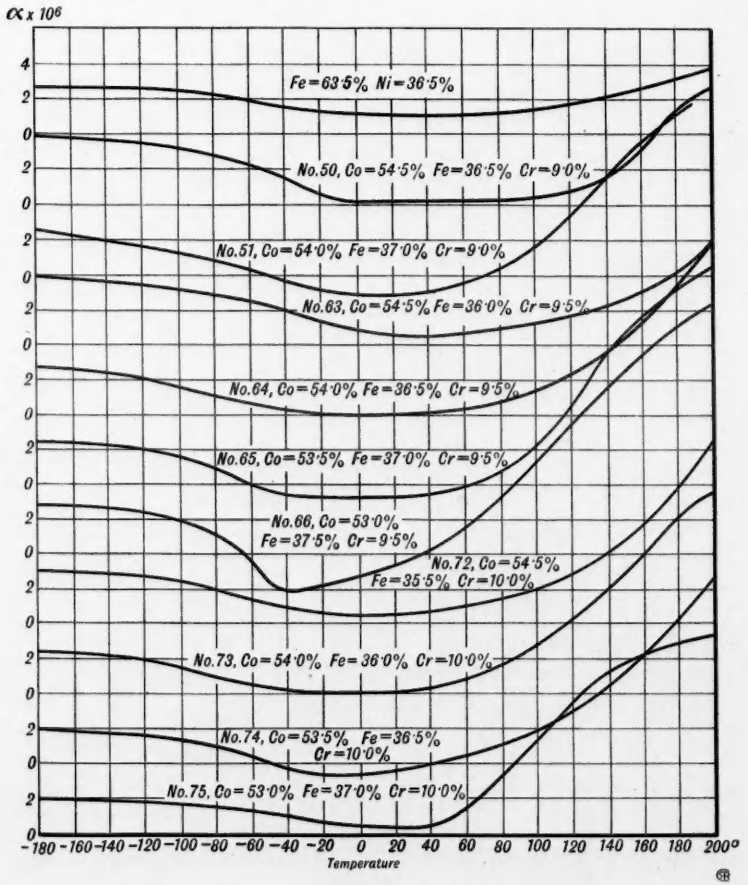


FIG. 2—*Thermal Expansion of Alloys of Cobalt, Iron and Chromium.*

increases. Further study of the alloys having compositions within the zone of small expansibility showed that certain alloys contracted over a certain range on heating in the vicinity of room temperature; these alloys also gave no allotropic transformation corresponding to A_3, even when cooled to $-252{\cdot}7$ deg. Cent., a behaviour different from that

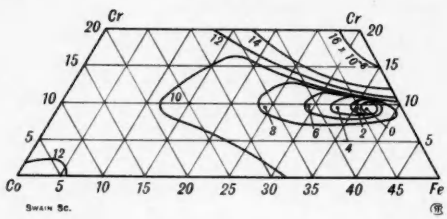


FIG. 3—*Coefficients of Expansion of Co-Fe-Cr Alloys.*

of "Super Invar." The variation of coefficient of expansion of a number of alloys (including "Invar"), with change of temperature, is shown in Fig. 2, and the coefficients at 20 deg. Cent. in Table I.

TABLE I.—*Expansibilities of Alloys of Compositions Within Zone of Low Expansibility.*

Alloy number.	Cobalt, per cent.	Iron, per cent.	Chromium, per cent.	Nickel, per cent.	Coefficient of expansion (20 deg.).
	—	63·5	—	36·5	$1{\cdot}2\times10^{-6}$
50	54·5	36·5	9·0	—	0·2
51	54·0	37·0	9·0	—	−1·2
63	54·5	36·0	9·5	—	0·5
64	54·0	36·5	9·5	—	0·0
65	53·5	37·0	9·5	—	−0·8
66	53·0	37·5	9·5	—	−0·6
72	54·5	35·5	10·0	—	0·5
73	54·0	36·0	10·0	—	0·0
74	53·5	36·5	10·0	—	−0·5
75	53·0	37·0	10·0	—	0·2

These alloys were found to resist corrosion in salt solution, and were named "Stainless Invar" (Cobalt 54·0, Fe 36·5, Cr 9·5 per cent.). Magnetic tests showed expansion (magnetostriction) up to a field intensity of 43 gausses, a contraction from 43 to 470 gausses, and a further expansion at higher intensities. The electrical resistance at 20 deg. Cent. was $66{\cdot}6\times10^{-6}$ and the temperature coefficient between 0 deg. and 40 deg. Cent. was $0{\cdot}832\times10^{-3}$.

The discovery of an alloy superior to "Invar" as regards thermal expansibility and resistance to corrosion is of considerable value, and may lead to new applications of such materials, as well as to the more extended study of the properties of cobalt-rich alloys.

Letters to the Editor.

(We do not hold ourselves responsible for the opinions of our correspondents.)

TUBE CORROSION AND OIL.

SIR,—Mr. Bullen's interesting article in your issue of December 28th, 1934, on the corrosion set up in condenser and cooling tubes by drops or patches of oil, provides practical examples of a phenomenon which was demonstrated in the laboratory some years ago. In a paper contributed, by invitation, to the American Oil Corrosion Symposium held in 1926, I showed* that drops of an inert oil resting against a metallic surface increased the chance of attack commencing at susceptible points in the annular crevice between oil and metal, and often led to a ring of corrosion. This was demonstrated, with several metals, several watery liquids, and several "oils," the result being shown most clearly by "oils" heavier than water (*e.g.*, carbon tetrachloride), which produced the results on the *upper* side of a metal surface, although it was also obtainable on the *lower* sides as using oils lighter than water. The phenomenon was not attributable to any corrosive constituent of the oil, but was due to the cranny formed around the point of contact between the oil globule and metal; it could equally well be produced by contact with a rounded glass surface.† Two explanations appeared to merit consideration, namely:—

(*a*) *Differential Aeration.*—The cranny, less accessible to diffusing oxygen than the rest of the surface, is liable to be anodic to the rest, and to suffer intensive attack.

(*b*) *Rival Interface Action.*—The product of incipient corrosion, which would in normal circumstances remain in physical contact with the metallic surface, smothering the attack, will, at the point in question, adhere to the oil, glass, or other foreign body, so that attack will not be smothered.

In the case of corrosion set up by contact with glass, statistical experiments carried out recently by R. B. Mears‡ in this laboratory, indicate that the first mechanism is much the most important cause of trouble. In the case of corrosion set up by inert oil, also, the differential aeration explanation is the one favoured both by Mr. Bullen and myself.

It may well be asked why the trouble is not met with more frequently, seeing that in many parts of the sea oil has become almost an essential constituent of the water. However, as I pointed out in 1926, the conditions favourable to this type of attack will not often arise, and subsequent work§ on "corrosion probability" makes it clear that actually they will be highly exceptional. To start cranny attack, it is necessary that a susceptible spot on the metal shall chance to fall within the cranny area formed by the edges of the oil drop. Furthermore, the attack will only be intense as long as the anodic area remains very small. If loose corrosion product accumulates, it will spread out and become less intense. If the water flowing through the tube carries off loose products, the attack may remain localised and intense, but a rapid water flow is liable to carry off—or even in time dissolve—the oil drop. Thus the range of flow rate favourable for serious attack may well be narrow. Many practical cases of corrosion by oil patches appear to be connected with the presence of mud or silt along the oil-water interface, which has kept stabilised the position of the oil patch on the metal—a matter mentioned also by Mr. Bullen. Mr. Bullen's explanation of a cause of wastage, which is likely to assume increasing practical importance, brings the phenomenon into line with many well-substantiated cases of corrosion of metal immersed in water or buried in soil.

ULICK R. EVANS.

The Metallurgical Laboratories,
Cambridge University,
January 23rd.

* U. R. Evans, "Petroleum Development and Technology in 1926," page 469 (Amer. Inst. Min. Met. Eng., 1927).

† U. R. Evans, "Corrosion of Metals," 1926 edition, p. 91 (Arnold).

‡ R. B. Mears and U. R. Evans, "Trans.," Faraday Soc., 1934, 30, 417.

§ R. B. Mears and U. R. Evans, "Trans.," Faraday Soc 1935, 31 (paper now in the Press).